SHAPE MEMORY EFFECT CHARACTERIZATION OF MODIFIED BETA-BRASS ALLOYS

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THESIS

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of

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ABSTRACT

Modified beta-brass alloys are tested at various temperatures to investigate the pseudoelastic and memory properties of the alloys. The effects of thermomechanical cycling, and prestrain on the beta-brass alloys are determined. A stress-strain relationship for the alloys at various test temperatures is developed.

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I. INTRODUCTION

The shape memory effect, whereby a material apparently plastically deformed reverts to its predeformed shape upon heating, has received wide attention in recent years. The shape memory phenomena was first observed at the United States Naval Ordnance Laboratory in about 1962. The original discovery was made in equiatomic TiNi and since then much research has been applied in understanding the mechanism of the memory effect and in applying materials which exhibit this phenomena to practical use. The shape memory effect has not been limited to TiNi alloys but has also been recently discovered in other alloy systems [Ref. 5].

The characteristics of the memory effect in these other alloy systems have not been completely determined as yet. Much work has been done in terms of phenomenological crystallographic studies of the memory effects [Refs. 5 and 7], but characterization of the mechanical features of the memory effect in alloy systems other than TiNi has not received widespread attention. To make practical application of the memory effect, the determination and characterization of its properties is important. Such features of the memory effect as the stress of reverse transformation, critical temperatures of the transformation and the effect of alloying and prestrain play an important role in practical applications. Some work has been done using beta-brass alloys. Pops [Ref. 6] has worked with these alloys in describing the pseudoelastic effect. Eisenwasser and Brown [Ref. 1] have also worked with these alloys to characterize the memory effect. Most of their work has dealt with the effect of deformation temperature on the alloy. Much more work is needed in this

and also in areas which have previously been mentioned to give a mechanical characterization of the alloy.

II. THEORY

Much study has been done in an attempt to determine the mechanism of the memory effect. According to previous work in determining the mechanism of the memory effect it appears that there are a number of essential characteristics for the effect to be present. Alloys possessing this effect appear to:

1. have a thermoelastic martensitic transformation;
2. have an ordered atomic arrangement in the parent and martensitic phases;
3. have internal twins in the martensite.

According to Otsuka and Shimizu [Ref. 4] the origin of the memory effect lies in the mobile property of the boundaries between martensite and matrix, and between twins, and in the coherent nature of these boundaries. A study of the memory effect in different alloy systems has shown that the alloys often have a bcc structure at high temperatures, which on quenching undergoes a low-hysteresis martensitic transformation. The memory effect can occur variously on deformation of an alloy either (i) entirely in the martensitic state ($T < M_f$) (where T is the deformation temperature), (ii) partially martensitic ($M_f < T < M_s$), or (iii) in a temperature range above M_s where deformation induced martensite, once formed, is stable ($M_s < T < M_d$). Deformation causes two basic interface reorientations: those involving (i) M:A interfaces and (ii) M:M interfaces (where M is the martensitic phase and A is the parent or "austenitic" phase). The unique feature is that when subsequently

heated the martensite reverts in such a way that the apparently plastically deformed specimen reverts to its original shape.

Any specimen which is stressed undergoes an elastic deformation until the yield point is reached. It then has a combination of elastic and plastic deformation. Classical elastic deformation is totally recoverable on unstressing, while classical plastic deformation involves slip via dislocation motion, which is not recoverable. In the present case, we must also consider modes of deformation such as twinning and stress-induced martensitic transformation, and their respective reversion behavior. In a shape-memory effect alloy, three separate processes are involved in recovery of a stressed specimen. The first process involves a total recovery of the classical elastic deformation as mentioned above. A second process has been described as the superelastic (pseudoelastic) effect and involves an immediate martensitic reversion upon removing the applied stress. This reflects the feature of martensite formation as an additional mode of deformation. The third process involves thermally activated martensitic reversion or "memory" of the alloy. The reversion of a martensite may be to either:

(i) the parent (high temperature) phase;

(ii) a (substructurally) different martensitic phase;

(iii) a different distribution of the martensite structural units (plates) in the microstructure.

In analyzing the results of the present experimental data the elastic recovery can be accounted for but the superelastic and thermally activated martensite reversion cannot be separated in all cases.

The temperature dependence of the memory effect has been defined in Cu-Zn based alloys by Eisenwasser and Brown [Ref. 1]. The critical

temperatures are A_s , A_f , M_s , M_d , and M_f . On cooling, martensite begins to form from austenite at the M_s temperature and continues to form until the M_f temperature is reached. At this temperature the specimen is 100 percent martensite. On heating the formation of austenite does not begin until the A_s temperature is reached and is completed at the A_f temperature. For specimens compression tested above A_f , the strain recovery is due largely to the pseudoelastic effect. Strain recovery of specimens tested between A_s and A_f is due to a combination of the memory and pseudoelastic effects. Below M_s recovery of specimens is largely a result of the memory effect. The M_d temperature corresponds to the appearance of strain-induced martensite in alloys. This temperature is higher than M_s and is the temperature at which martensite can first be seen in an alloy. Table I is a listing of the critical temperatures of the alloys tested. The range for the critical temperatures is due to the variance in composition of specimens of the same alloy which resulted from increased loss of zinc with each heat treatment.

The thermoelastic nature of the memory effect implies that the thermal and mechanical behavior of the material are interrelated. The martensitic transformation of the bcc parent phase is strain sensitive. In the case of "thermoelastic" martensite, the martensitic plates grow continuously as the temperature is lowered and shrink on heating. The rate of growth of the martensitic plates appears to be solely a function of the rate of change of temperature, thus being thermoelastic in nature.

The ordered arrangement of the parent and martensitic phase is important because of the reverse transformation which takes place on heating. When an atomically ordered alloy transforms martensitically, the martensite reverts as a unit to the original austenitic orientation. This behavior

is not applicable to disordered alloys in which, on reversion, a number of austenitic orientations are produced from a single martensitic orientation.

The shape memory effect, because of the reversion which takes place, can be deduced to not require dislocation motion as its mode of lattice invariant deformation. Deformation of a material by slip is by nature an irreversible process and would not be consistent with a material able to revert repeatedly to its original shape. A martensitic material which possesses the shape memory effect therefore would not be expected to require much dislocation generation or motion during transformation. As pointed out by Wayman and Shimizu [Ref. 9], memory martensites usually contain internal twins as substructure, indicating that twinning is the mode of lattice invariant deformation. The twinning process results in a plastic deformation process whereby one twin grows at the expense of another within individual martensite plates. Slip does not occur and thus the reversion can take place. Therefore twinning and detwinning rather than slip is the mode of substructural deformation in the martensites of alloys which exhibit the memory effect. Another mechanism of deformation for alloys possessing the strain memory effect was proposed by J. D. Eisenwasser and L. C. Brown [Ref. 1]. A martensite forms on cooling with negligible dimensional changes in the specimen (< 0.1 pct) provided it is under zero stress. This is argued on the basis that several orientation variants of the martensite form in the matrix, and on deformation any elongation associated with one particular variant is cancelled out by shrinking of the others. This is, on stressing one variant grows at the expense of the others; in the ultimate one martensitic orientation would result. On release of the stress only the elastic strain is released.

Only on heating between A_s and A_f, to where the beta matrix exists is all strain recovered. This explanation differs from that in which the memory effect is attributed to the twinning process. Two possible interfaces could result from this process. One is between the martensite formed upon cooling and the stress induced martensite (M:M'). The second type of interface would be between the martensite and the beta matrix $^{(M:\beta)}$. The first type would be present if the specimen were (i) fully martensitic (below M_f), (ii) partially martensitic (between M_s and M_f), or (iii) non-martensitic (above M_s but below M_d). The second type of interface is present in only situations (ii) and (iii) of the above.

III. EXPERIMENTAL PROCEDURE

A. ALLOY PREPARATION

All alloys were prepared from metals of 99.999 percent purity. The materials were cleaned and pickled so as to limit the amount of grease and oxide present. The metals were weighed on a Sartorious Balance and placed in fused silica capsules. Before sealing, these capsules were evacuated with a mechanical pump, baked at a temperature of approximately 200°C to remove absorbed water and then backfilled with a half atmosphere of helium. The alloys were then heated to approximately 1100°C for a period of twenty minutes. During this twenty minute period the molten alloy was shaken vigorously at five minute intervals to assure mixing of the metals. The alloys were then quenched by immersing the capsules in water. See Appendix A for further details on the melting procedures.

The alloys were then resealed in silica capsules (using the same sealing procedure) for homogenization. The alloys were heated to a temperature of 825°C and left at that temperature for 48 hours. On

removing the alloys from the furnace they were quenched in an iced brine solution. This process was used in order to obtain a compositionally homogenous specimen. Upon completion of this process the alloys were reweighed to determine the percentage of weight lost during the process. The alloys were then machined into specimens 0.20 inch in diameter and 0.30 inch in height to be used for compression testing. Fig. 1 is an enlarged picture of a typical specimen.

After machining, the alloys were resealed in quartz capsules and solution-treated at 825°C for one hour and then quenched in an iced brine solution. The alloys were then electropolished and examined to insure a beta phase alloy existed. Those alloys which had large amounts of alpha phase were resealed and resolutioned for twenty-five minutes.

Three Cu-Zn base alloys, compositionally modified by small amounts of Sn, were prepared for the experimental program. The three alloys and their compositions are given in Table I. The initial structure of each specimen was photographed at 100X prior to testing (see Figs. 2, 3, 4 and 5).

TABLE I

ALLOY COMPOSITION AND CRITICAL TEMPERATURES

ALLOY	% Cu	% Zn	% Sn	A_f °C	A_s °C	M_s °C	M_f °C
4	64.9	32.9	2.2	15, 40	0, 25	0, 25	-15, 10
5	63.8	34.55	1.65	-35, -10	-50, -25	-50, -25	-65, -40
6	62.9	35.9	1.2	-85, -60	-100, -75	-100, -75	-115, -90

The first temperatures in columns five through eight are the nominal temperatures of the alloys. The second temperatures are the estimated temperatures due to increased zinc loss with each heat treatment.

Figure 1. Enlarged View of a Typical Specimen Showing .2 Inch Diameter and .3 Inch Height.

Figure 3. Microstructure of a Specimen Having an Almost Complete Beta Phase (500X).

Figure 3. Microstructure of a Specimen Having Large Amounts of Alpha Phase Widmenstätten Present (650X).

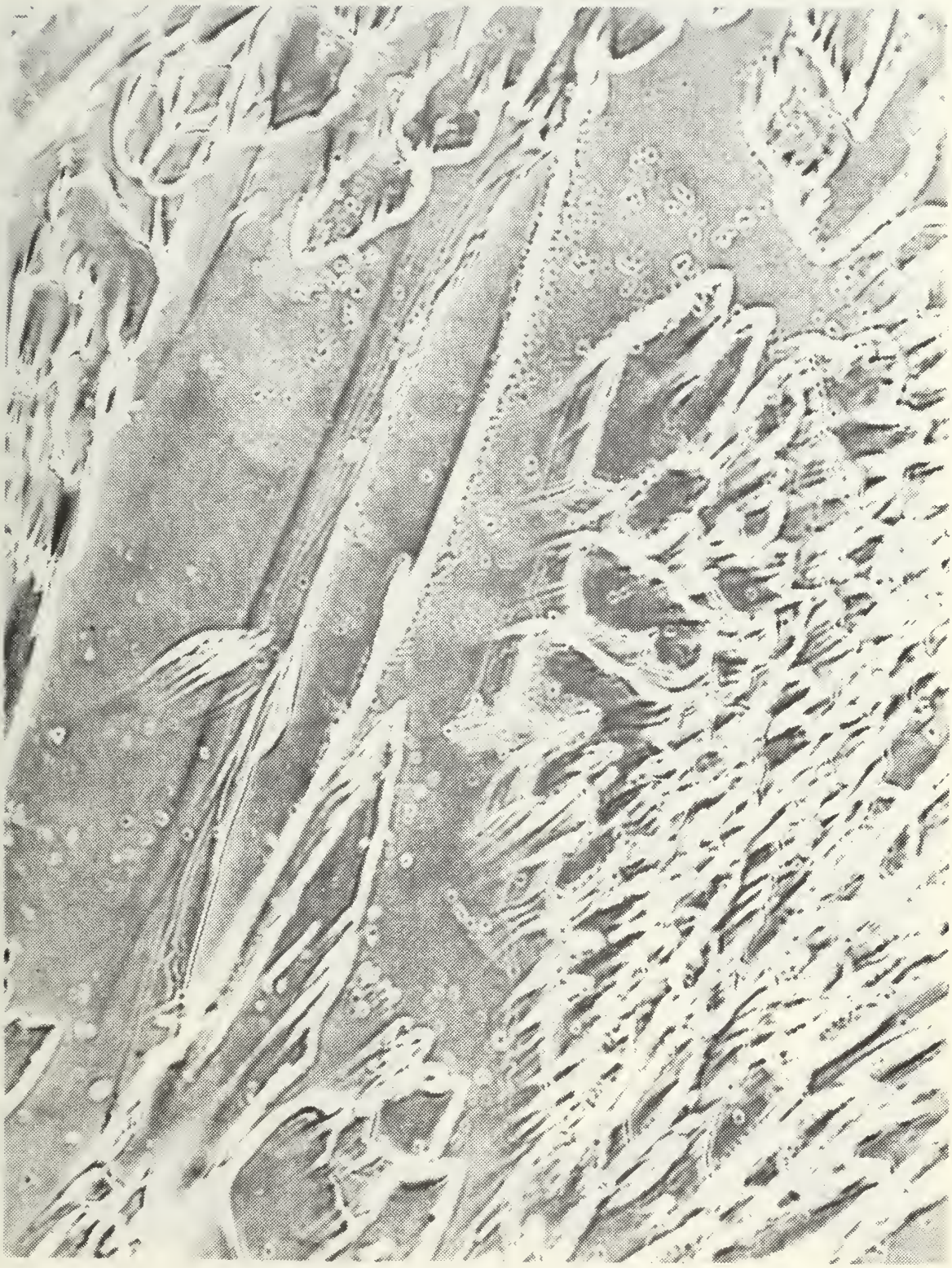

Figure 4. Microstructure of a Two Phase Specimen (650X).

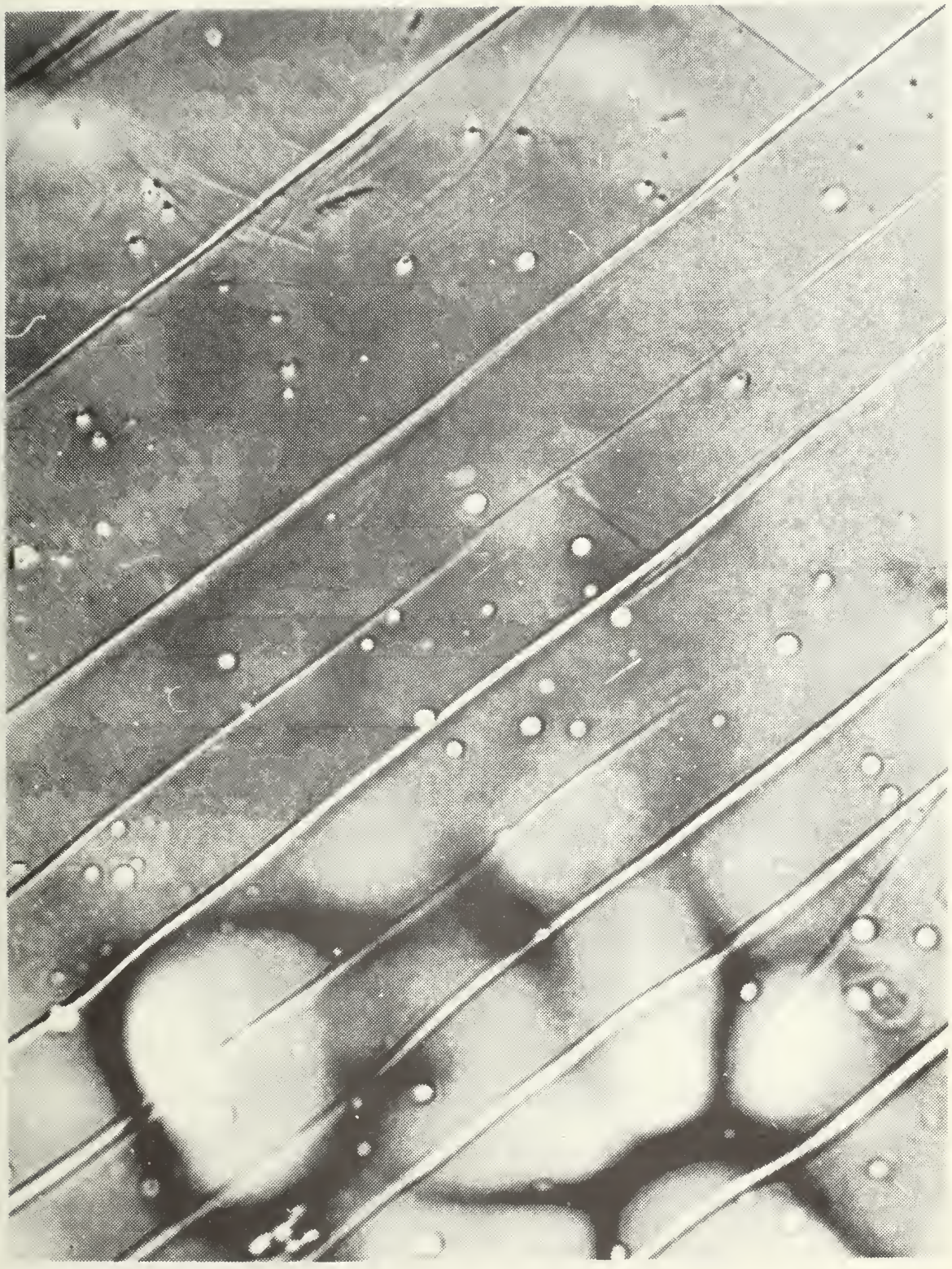

Figure 5. Microstructure of a Largely Beta Phase Specimen With Thermal Martensite Present (650X).

B. MECHANICAL TESTING

The initial mechanical testing investigated the stress-strain behavior of the beta-brass alloys at several temperatures. The test temperatures were ambient (25°C), -75°C (obtained by putting dry ice in a dewar with alcohol), and -196°C (the temperature of liquid nitrogen).

Subsequent testing investigated the effect of thermal and mechanical cycling on the different alloys. One form of cycling involved the loading of a specimen to a given load (approximately 9000 psi) at a reduced temperature. The specimen after loading was heated to ambient temperature to measure the increase in stress caused by reversion of the martensite. A measurement of specimen length was made after unloading and the procedure repeated. The specimens were also cycled at the reduced temperatures with an increasing load applied. The test in this case was performed in the same manner as the constant stress cycling but with increased stress applied to the specimen with each cycle. A limited number of tests were also performed to determine the effect of prestrain on the alloys. To do this a load was applied to a specimen at 25°C and then unloaded. The specimen was then reduced in temperature and reloaded. The temperature was then raised to ambient temperature and a measure of the amount of recovery made.

To test the specimens in the manner described a compression cage mechanism as shown in Figs. 6 and 7 was designed. The cage was made of stainless steel with three guide bars. The center bar which was connected from the upper moving plate to the load cell had the same diameter as the three guide bars. This was done to help to insure a uniform heat transfer when this portion was placed in the reduced temperature. Just above the reduced temperature medium the center bar was given a reduced diameter to

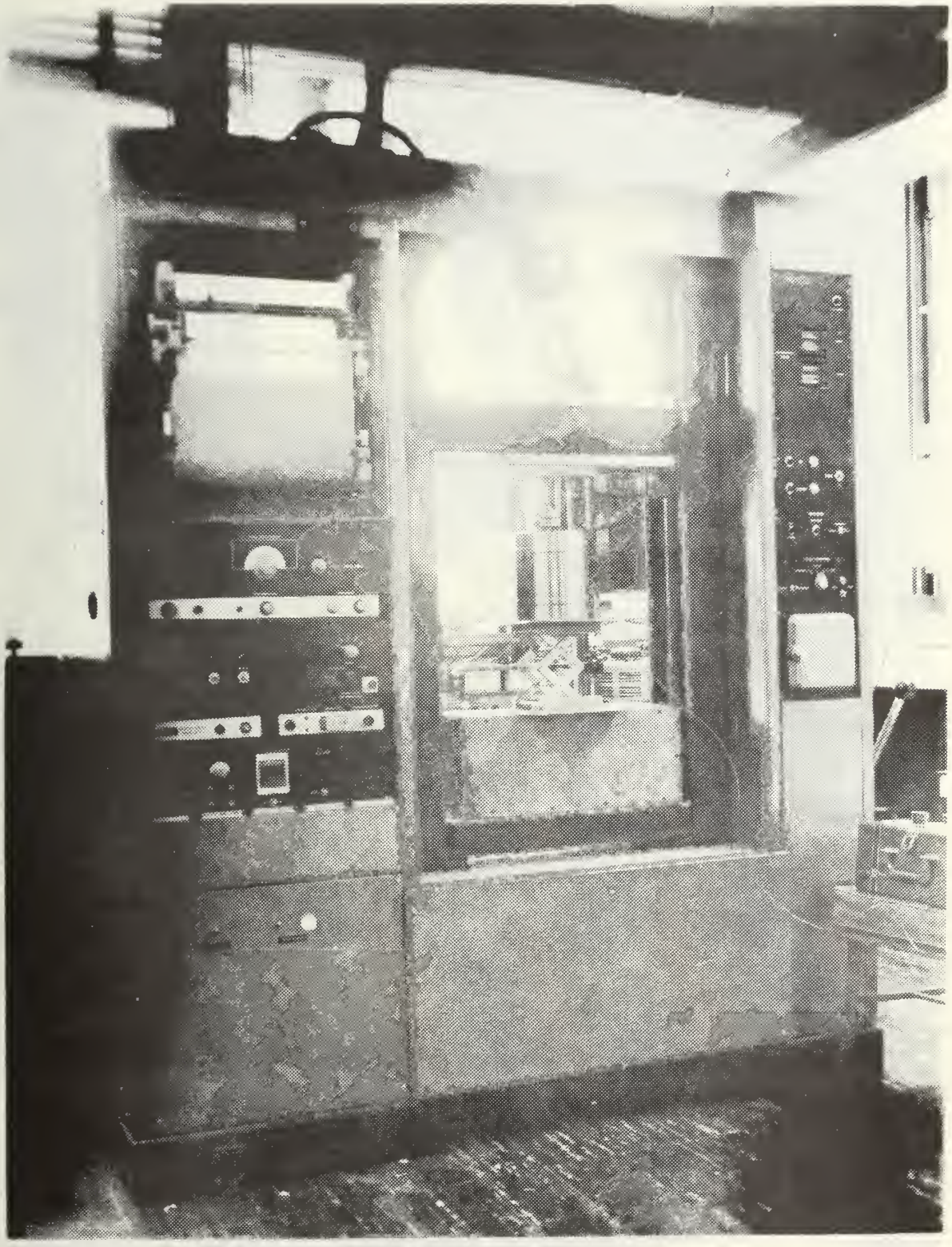

Figure 6. Instron Testing Machine With Compression Cage.

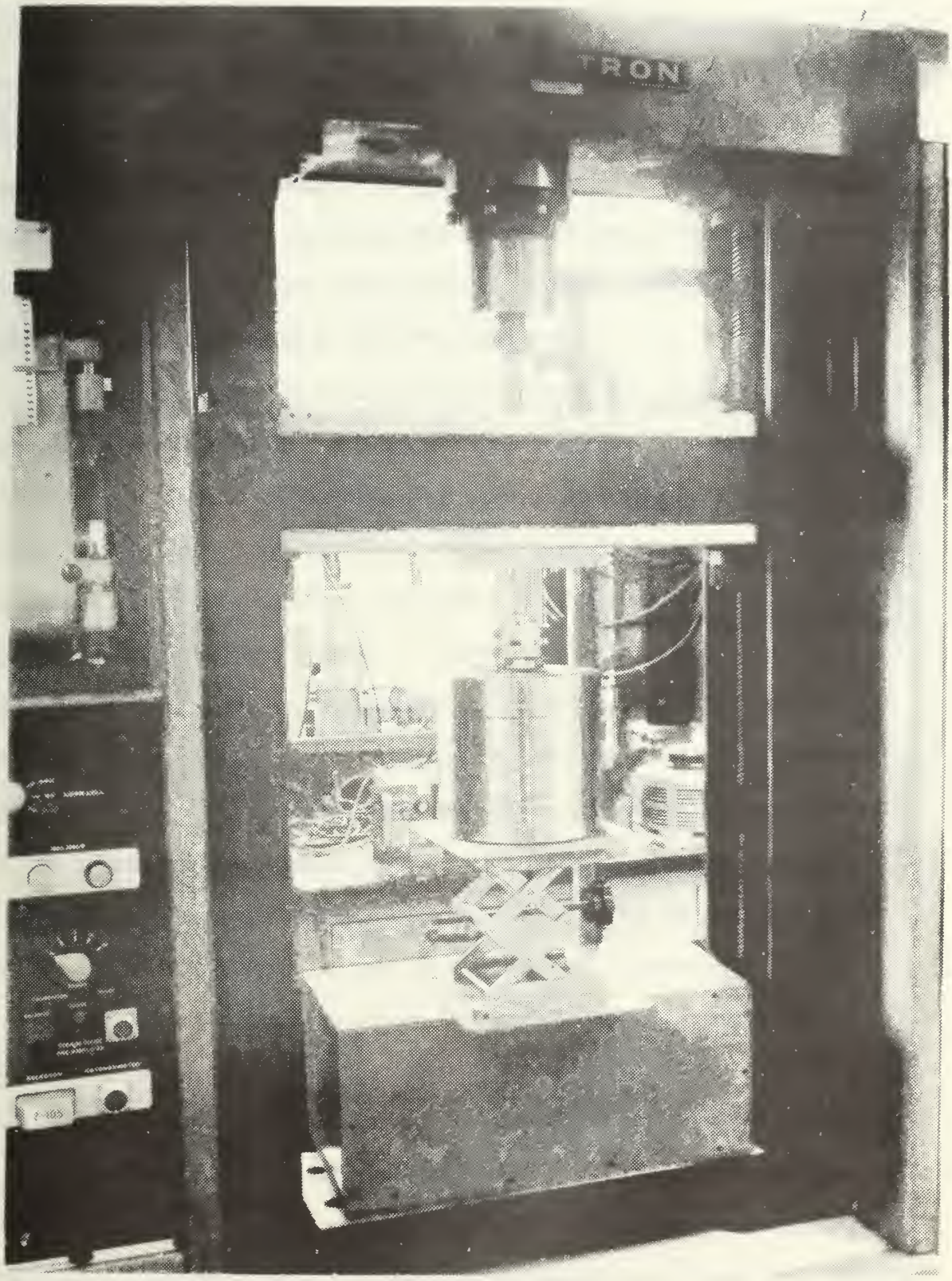

Figure 7. Load Cell and Compression Cage Attachment to Instron.

provide a limited heat transfer path between the low temperature region and the remainder of the bar. The GR-tension-compression load cell of the Instron testing machine was attached as shown in Fig. 7. The cell was moved to this position to prevent any slipage due to the attaching bolts being placed in tension in the original setup. To calibrate the load cell for testing, the electronic calibration was used after a check was made by hanging weights from the load cell.

To insure as accurate measurement as possible a "fingerprint" of the cage apparatus was taken for each test procedure. These "fingerprints" were tests performed on two tungsten carbide platens in exactly the same manner as the tests were performed on the various specimens. It was necessary to do this because there was a heat transfer along the load carrying bar of the mechanism. As a result of this heat transfer there was a continuing contraction of the bar as the temperature was lowered followed by an elongation on heating. A test was also run with the specimen present but no load applied. After lowering the temperature of the mechanism the platens were moved to a position just touching the specimen. The temperature was then raised to 25°C; no increase in load due to relative specimen or cage elongation was indicated.

A measurement of each specimen was made with a micrometer prior to testing. The specimen was placed between two tungsten carbide platens with a piece of teflon tape separating the specimen from the platens to minimize barreling. The specimen mounted as such was then centered between the two plates of the cage mechanism to undergo a given test. After each loading given to a specimen a length measurement was made with the micrometer. In performing tests at the reduced temperatures of -75°C and -196°C, it was discovered that removing the low temperature atmosphere

caused icing on the cage, which interferred with smooth motion. To eliminate this, the cage mechanism and specimen were thereafter immersed in alcohol at ambient temperature immediately following removal of the cooling medium. This prevented the icing and also provided more rapid heat transfer from the warm alcohol to the cold mechanism. As the temperature of the alcohol was lowered, it was replaced by warmer alcohol. To obtain stress-strain curves from the load-deflection curves which resulted from testing of the specimens, the effects of the non-rigidity of the cage had to be eliminated. This was accomplished by subtracting the "fingerprints" of the cage without the specimens from the curves of the alloys. This resulted in a load-elongation curve for the specimen being tested which could be converted to a stress-strain curve representing the alloy behavior alone.

IV. RESULTS AND DISCUSSION

A. STRESS-STRAIN RELATIONSHIPS

The stress-strain relationship for the three compositionally modified Cu-Zn base alloys are shown in Figs. 8, 9 and 10. A wide variance in the curve shapes at the three temperatures is readily noticeable. The stress-strain curves of the three alloys at 25°C have a shape which is typical of most polycrystalline metals and alloys. Each alloy exhibits a linear or elastic portion followed by a plastic region. In this set of curves a distinct yield point is present in each case. The two sets of low temperature stress-strain curves differ noticeably from the 25° curves. The linear elastic coincident region of the 25°C curves is not present at the reduced temperatures. The stress-strain curves for the alloys at low temperatures exhibits properties similar to those of the Cu-Zn-Sn

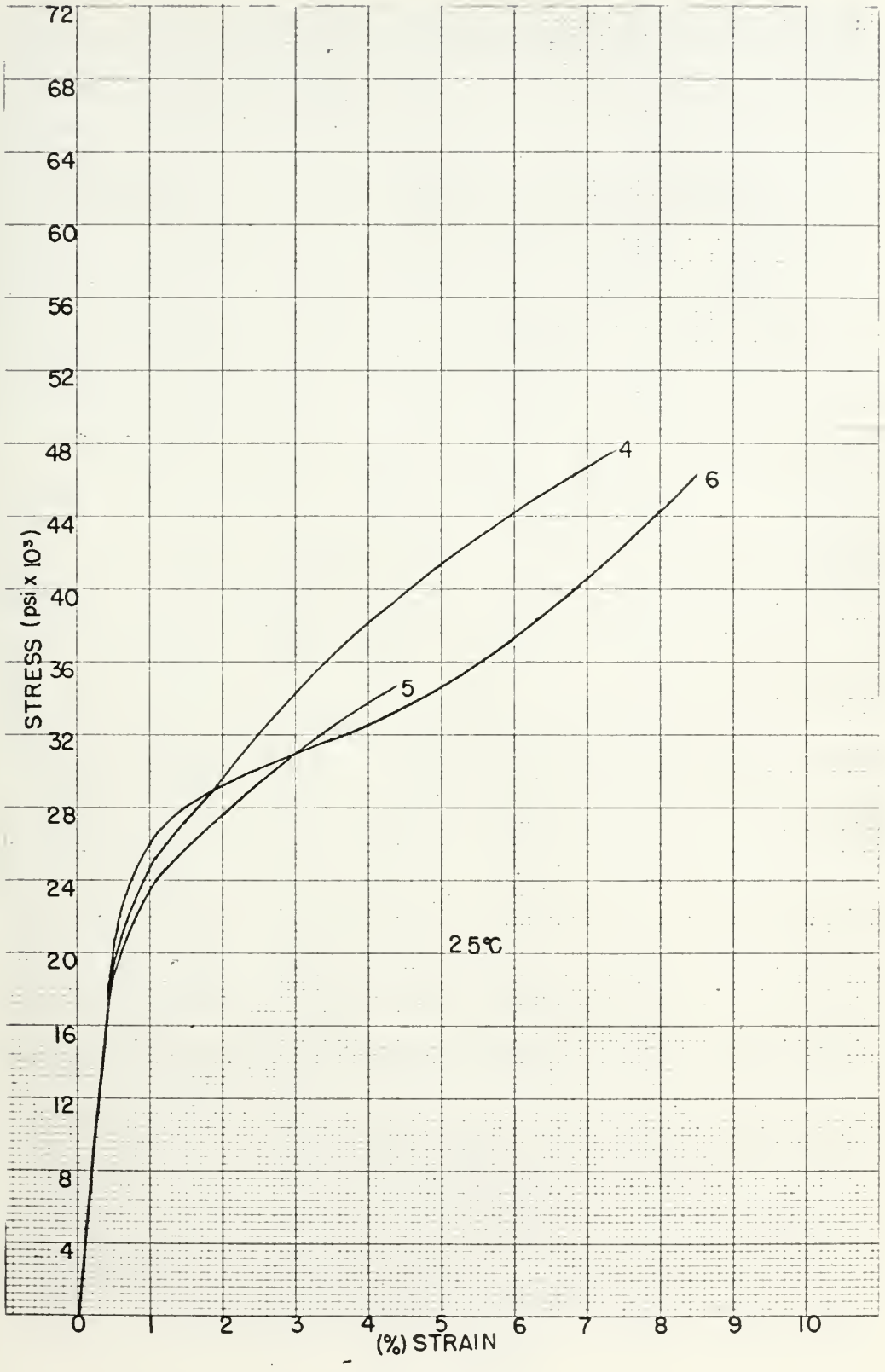

Figure 8. 25°C Stress-Strain Curves for the Three Alloys Tested.

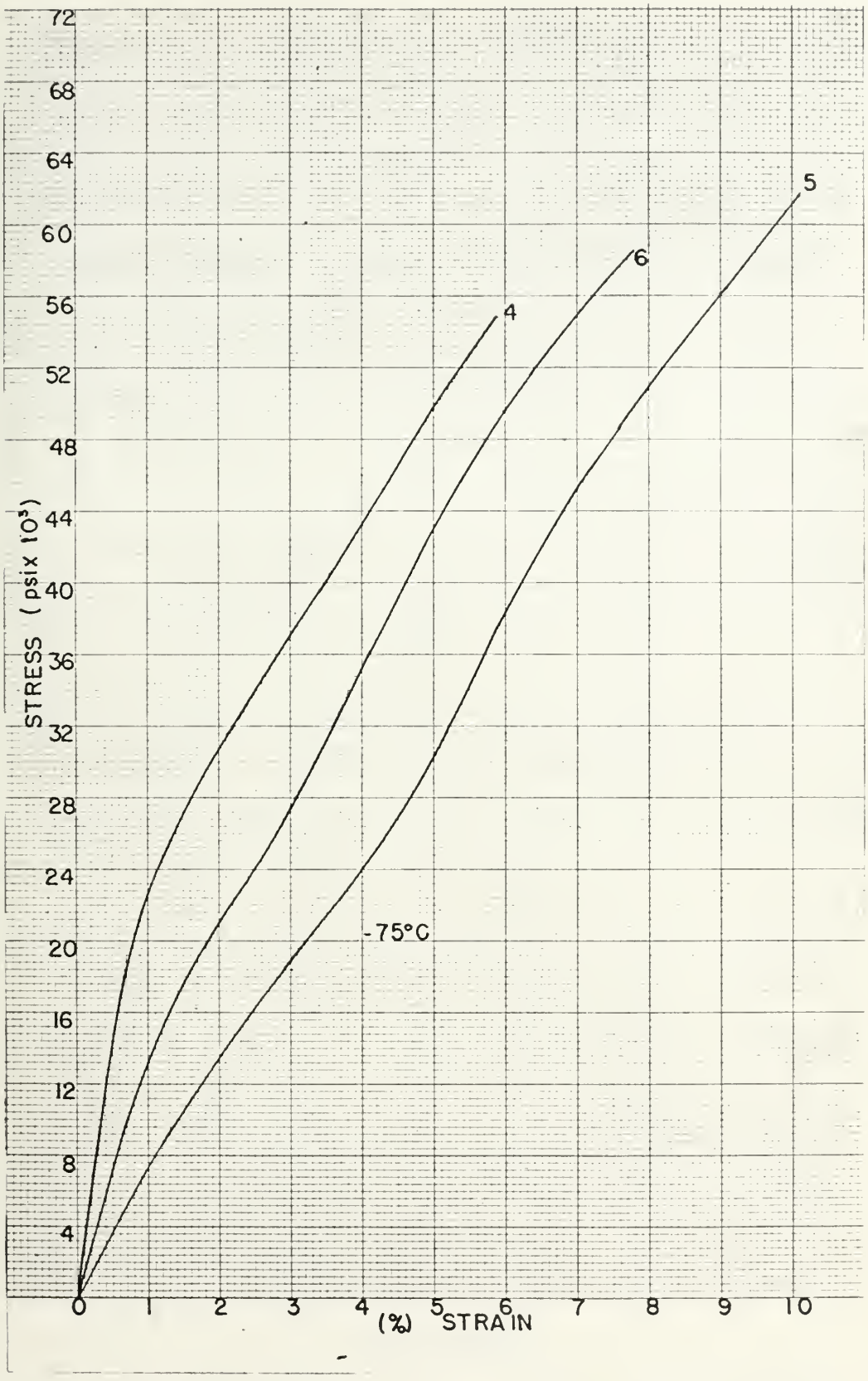

Figure 9. -75°C Stress-Strain Curves for the Three Alloys Tested.

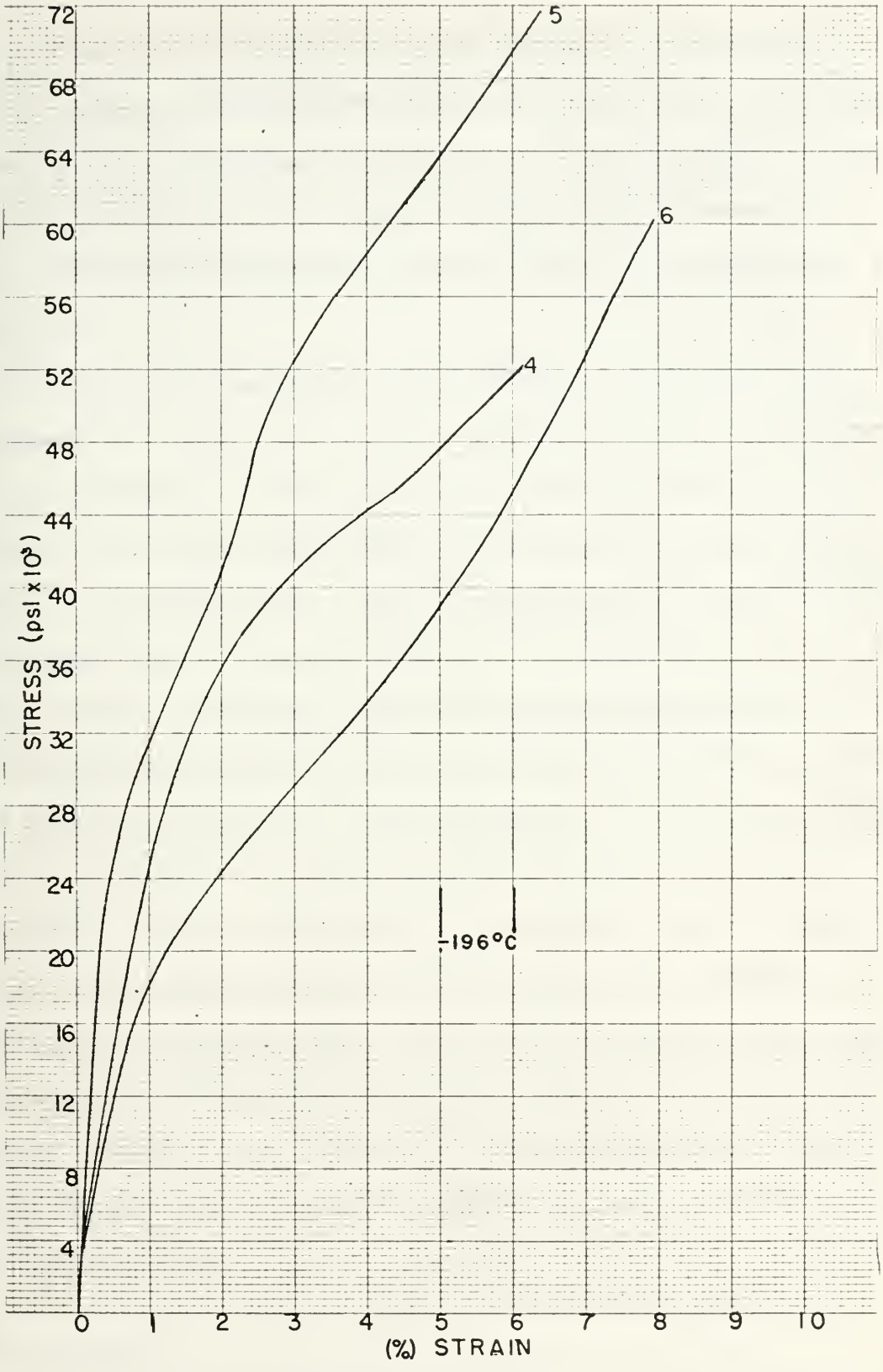

Figure 10. -196°C Stress-Strain Curves for the Three Alloys Tested.

alloy investigated by Eisenwasser and Brown [Ref. 1]. It is noticed that there is a decrease in what would normally be considered the yield stress for the alloys with decreasing temperature. It is also noticed that the rate of strain hardening for the alloys decreases as the temperature increases. Both these observations were also seen by Eisenwasser and Brown [Ref. 1].

In obtaining the stress-strain relationships for the three alloys, it was noticed that there was a variance in the data for the same alloy at the same temperature. This can be seen by comparing initial stress-strain data of the cycling experiments with the stress-strain data of these curves. A primary cause of this phenomena was the large grain size of the specimens which in many cases approached the diameter of the specimen (.2 inches). The large grains of the specimens would mean a "quasi-single crystal" and thus the mechanical behavior of the specimen would be orientation dependent. Another explanation for the differences of the stress-strain data is that the amount of Widmanstätten alpha secondary phase, which is softer than the beta phase, varied between specimens. A third reason for the variance could be the different homogenizing heat treatments given to specimens of the same alloy. With each heat treatment to remove the alpha phase present there was an additional loss in zinc via volatilization. X-ray diffraction and quantitative metallography could be done to correlate the behavior of the specimens with its microstructure.

B. CYCLIC TESTING

The results of cyclic testing of the various alloys are presented in Tables II through VI. Table II is a listing of the data obtained from

25°C cycling with an increasing load on each cycle. Alloy 4 at this temperature is approximately at its M_s temperature while 5 and 6 are both above their A_f temperatures. The recovery exhibited by the three alloys at this temperature is due largely to the pseudoelastic effect. There is no memory present at this temperature because the alloys have not been thermally activated by heating above A_f. It can be seen from examining this table that at low strains 100 percent recovery is exhibited by the specimens. The percent recovery in general decreases with increasing strains. It appears that the decrease in strain recovery is slow until approximately four percent strain is applied. At greater strains, the strain recovered on unloading decreases rapidly with increasing strain until just elastic recovery occurs.

Table III is a list of the results of increased load cycling at -75°C (data was obtained at this temperature for alloy's 5 and 6 only), and Table IV is the cyclic loading data from testing at -196°C. At a temperature of -75°C alloy 5 is below its M_f temperature while alloy 6 is approximately at its M_s temperature. At this temperature the amount of recovery in alloy 5 would be expected to be largely dependent on the memory effect (thermally activated reversion) while that of alloy 6 would be partially memory effect and partially pseudoelastic recovery (immediate reversion). The results as shown in Table III show that at -75°C there appears to be a limit to the percent of strain recovery as was also seen at 25°C. At this temperature a measure was also made of the increase of stress as a result of reversion. The increase of stress also appears to reach a limit with increasing strain, after which it decreases.

Table IV is a summary of the results of cyclic loading of alloys 4, 5 and 6 at -196°C. All three alloys are below their respective M_f

temperatures at -196°C and the recovery exhibited is expected to be due only to the memory effect (and the usual elastic recovery of the specimen) with minimal pseudoelastic effect. At this temperature the limit of percent recovery and increase in stress with reversion is also noticeable.

The three tables show that the amount of strain recovered by these three alloys is approximately five to ten times greater than the elastic recovery of a 60-40 brass alloy. At a load of 60,000 psi a Muntz Metal can be expected to recover approximately .4 percent strain. From the alloys tested it was seen that alloy 6 recovered a maximum of 3.96 percent strain or ten times as much. The three tables also show that a limit in applied strain does appear to exist beyond which there is little recovery of the specimen and that for small strains one to two percent a recovery of 80 to 100 percent is obtainable. With increasing plastic deformation, the percent of strain recovery decreases. The results obtained also show that the alloying of the specimen has an effect on the recovery it exhibits. The effect of alloying on recovery cannot be determined directly by analyzing the percent recovery for a given strain. One can see by looking closely that in all cases the stress of reversion for a given loading is higher for alloy 6 than 5 and also higher for 5 than 4. It is also shown that for a given permanent deformation given to the specimen the percent of recovery is also highest for 6, with alloy 5 again greater than 4. This leads to the conclusion that the alloying has an effect on the recovery of the specimen and that the greater the percent zinc in the beta phase, the higher the recovery at a given temperature.

The results of constant load cycling at -75°C and -196°C are given in Tables V and VI respectively. The cycling in this case was done at relatively low loads and strain levels (a maximum strain of approximately

TABLE II

CYCLIC INCREASING LOAD DATA 25°C

25°C	STRESS (psi)	% STRAIN APPLIED	% PERMANENT STRAIN	% STRAIN RECOVERED	% TOTAL RECOVERY
4 - 7	4450	.3	0	.3	100
4 - 7	9500	1.45	.07	1.38	95
4 - 8	15900	4.2	1.095	3.105	74
4 - 8	28650	6.5	6.25	.25	4
	4775	.2	0	.2	100
	6350	.27	0	.27	100
	11800	.75	.07	.68	91
5 - 3	15900	.8	.13	.67	83.8
	20700	.85	.23	.62	73
	31850	1.8	.46	1.34	74.5
	41250	3.8	.96	2.84	75
	50900	4.2	2.20	1.92	46
	3975	.07	0	.07	100
	10350	.5	0	.5	100
	15900	.5	0	.5	100
6 - 1	25450	1.25	.27	.98	78.5
	36750	2.73	.6	2.13	78
	41250	3.63	.67	2.96	81

TABLE III

CYCLIC INCREASING LOAD DATA -75°C

-75°C	STRESS (psi)	% STRAIN APPLIED	% PERMANENT STRAIN	% STRAIN RECOVERED	% TOTAL RECOVERY	STRESS OF REVERSION
5 - 11	6350	.17	0	.17	100	1900
	12700	.47	.07	.40	85	160
	19100	.55	.07	.48	87.5	600
	25450	1.03	.17	.86	83.5	1300
	31850	2.78	.67	2.11	76	6100
	38200	3.87	1.34	2.53	65.5	5100
6 - 8	6350	.77	0	.77	100	6350
	12700	3.25	.20	3.05	94	15900
	19100	4.4	.54	3.86	88	14300
	25450	4.6	.70	3.90	85	10820
	31850	5.0	1.04	3.96	79	5400

TABLE IV

CYCLIC INCREASING LOAD DATA -196°C

-196°C	STRESS (psi)	% STRAIN APPLIED	% PERMENANT STRAIN	% STRAIN RECOVERED	% TOTAL RECOVERY	STRESS OF REVERSION
4 - 11	6350	.07	0	.07	100	600
	12700	.5	.17	.33	66	950
	19100	1.9	.76	1.14	60	2240
	25450	2.4	1.89	.51	21	3500
5 - 6	6350	.43	0	.43	100	1900
	12700	.6	0	.6	100	2700
	19100	2.0	0	2.0	100	7000
	25450	2.23	.26	1.97	89	9870
	31850	3.55	.41	3.14	88.5	6690
	38200	3.43	.52	2.89	77.5	4450
6 - 4	6350	.13	0	.13	100	1900
	12700	.7	0	.7	100	4100
	19100	1.3	.07	1.23	94.6	11000
	25450	1.53	.20	1.33	89	14300
	31850	1.87	.34	1.53	88.5	13850
	38200	2.5	.54	1.96	87.5	11780

TABLE V

CONSTANT LOAD CYCLING -75°C

-75°C 8750 psi	CYCLE	% STRAIN APPLIED	% STRAIN RECOVERED	% TOTAL RECOVERY	STRESS OF REVERSION
5 - 9	1	1.4	1.4	100	7310
	2	1.2	1.2	100	7480
	3	1.13	1.3	100	6840
	4	.9	.9	100	7640
6 - 7	1	1.27	1.17	92	8900
	2	1.47	1.30	89	11460
	3	1.5	1.26	84	11460
	4	1.67	1.33	80	12250

TABLE VI

CONSTANT LOAD CYCLING -196°C

-196°C 8750 psi	CYCLE	% STRAIN APPLIED	% TOTAL RECOVERY	STRESS OF REVERSION
4 - 9	1	-	100	2385
	2	-	100	2070
	3	.17	100	1590
	4	.30	100	1020
	5	.2	100	1900
5 - 5	1	.67	100	3320
	2	.77	100	2385
	3	-	100	3020
	4	.67	100	3650
	5	.4	100	2385
6 - 3	1	.67	100	2385
	2	.73	100	2240
	3	.4	100	3500
	4	.37	100	2540
	5	.53	100	2700

1.5% was investigated). Table V lists the results of cycling alloy 5 and 6 at -75°C. The results show that for approximately equal strains and loadings the stress of reversion of 6 is higher than 5. The percent recovery for alloy 5 is 100 percent while alloy 6 varies between 80 and 100 percent. Table VI tabulates the results for cycling at -196°C for all three alloys. There is little difference noted in the properties of the three alloys. All exhibit 100 percent recovery. The strain used for alloy 4 is less than that for 5 and 6 and its stress of reversion is less. Alloy 5 and 6 exhibit approximately the same results at this temperature and loading. The same results had been obtained for increasing load cycling of Table IV for the low loading of 6350 psi.

C. EFFECTS OF PRESTRAIN

In performing this test alloy 5 was loaded at 25°C to four percent strain. This same specimen was then loaded to approximately the same load at -196°C. The results of this test are presented in Table VII. The percent of recovery and the permanent deformation of the alloy are based on the length of the specimen after the 25°C prestrain and not the original length. The initial results at 25°C as shown are very similar to the results expressed in the cycling experiment for approximately equal loading. The results of the test at -196°C show that the strain recovery is less and the amount of permanent deformation greater for the same alloy after prestrain. The increase of stress due to reversion is approximately the same for this test as it was in the cycling test. From these results it appears that the amount of prestrain given to a specimen has a direct effect upon its recovery. Increased prestrain would be expected to cause a decrease in the ability of the specimen to recover with thermal activation.

TABLE VII

EFFECTS OF PRESTRAIN

TEMP °C	STRESS (psi)	% STRAIN APPLIED	% PERMANENT STRAIN	% STRAIN RECOVERED	% TOTAL RECOVERY
25 (pseudoelastic)	31990	4.03	.68	3.35	83
-196 (memory)	31830	3.50	.75	2.75	78.5

V. CONCLUSIONS

Alloys tested show marked pseudoelastic and memory effects.

Recovery of low strains (1 to 1.5 percent) is approximately total for the alloys tested.

Strains recovered for the alloys tested are from five to ten times greater than the normal elastic recovery of a regular 60 Cu - 40 Zn brass.

Alloying has a direct effect on the strain recovery of a specimen. Increased percent zinc of the beta phase alloy increases the recovery of the alloy. Maximum strain recovered was four percent for alloy 6.

The increase in stress due to reversion is alloy dependent. Increased percent zinc of the beta phase increases the stress of reversion. Alloy 6 had a maximum stress increase of 16,000 psi, alloy 5; 10,000 psi, and alloy 4; 3,500 psi. After this maximum is reached for each particular alloy, the stress of reversion decreases.

Prestrain caused a decrease in the memory recovery of the specimen tested.

APPENDIX A: INITIAL ALLOY PREPARATION

An attempt was initially made to cast copper zinc based alloys modified by aluminum instead of tin. Because of the possible reaction of aluminum with silica glass on heating to higher temperatures, it was necessary to place the alloying components in a graphite crucible before sealing in a quartz tube as shown in Fig.11. Several attempts were made to melt the components in this way, but each attempt to quench (in one case to shake and mix) the tube containing the molten alloy, resulted in an explosion. The explosions which took place on plunging the tube into a water bath were accompanied by an odor typical of rotten eggs. It was eventually decided that a sulfur-rich gas had been present within the tubes and that this subsequently reacted with hydrogen (derived from H_2O dissociated locally by the +1000°C heat of the tube) to form H_2S gas having this odor. The initial formation of the sulfurous gas in the tube can be attributed to impurities (probably sulfur or sulfur compounds) present in the porous graphite crucible. It is likely that on heating, these compounds were volatilized to form a gas which in turn embrittled the (initially high purity) quartz tube. The thermal shock of quenching then caused the fragile tube to explode, propelled by the high vapor pressure of the mentioned gas. Pure quartz glass normally sustains such thermal shocks extremely well. The argument proposed is reinforced by the observation that the graphite tubes if preheated (to volatilize and remove impurities) no longer gave rise to explosion of tubes into which they were sealed, heated, and quenched.

Figure 11. Graphite Crucible Containing Alloying Elements in a Silica Glass Tube.

REFERENCES

1. Eisenwasser, J. D. and Brown, L. C., "Pseudoelasticity and the Strain Memory Effect in Cu-Zn-Sn Alloys," Metallurgical Transactions, Vol. 3, p. 1359, June 1972.

2. Kushman, R. V. and Brown, L. C., "Pseudoelasticity and the Strain Memory Effect in an Ag-45 Pct. Cd Alloy," Metallurgical Transactions, Vol. 4, p. 423, February 1973.

3. Oishi, K. and Brown, L. C., "Stress Induced Martensite Formation in Cu-Al-Ni Alloys," Metallurgical Transactions, Vol. 2, p. 1971, July 1971.

4. Otsuka, K. and Shimizu, K., "Memory Effect and Thermoelastic Martensite Transformation in Cu-Al-Ni Alloys," Scripta Metallurgica, Vol. 4, p. 469, 1970.

5. Perkins, A.J., "Lattice Transformations Related to Unique Mechanical Effects," paper presented at Symposium on Phase Transformations in Less Common Metals: A Dialogue, Cleveland, Ohio, October 16-17, 1972.

6. Pops, H., "Stress Induced Pseudoelasticity in Ternary Cu-Zn Based Beta Prime Alloys," Metallurgical Transactions, Vol. 1, p. 251, January 1970.

7. Pops, H. and Massalski, T. B., "Thermoelastic and Burst Type Martensites in Copper-Zinc Beta Phase Alloys," Transactions of the Metallurgical Society of AIME., Vol. 230, p. 1662, December 1964.

8. Sastri, A. S. and Marcinkowski, M. J., "Deformation Behavior in the Near Equiatomic NiTi Alloys," Transactions of the Metallurgical Society of AIME, Vol. 242, p. 2393, December 1968.

9. Wayman, C. M. and Shimizu, K., "The Shape Memory (Marmen) Effect in Alloys," Metal Science Journal, Vol. 6, p. 175, 1972.

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ABSTRACT

Modified beta-brass alloys are tested at various temperatures to investigate the pseudoelastic and memory properties of the alloys. The effects of thermomechanical cycling, and prestrain of the beta-brass alloys are determined. A stress-strain relationship for the alloys at various test temperatures is developed.

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